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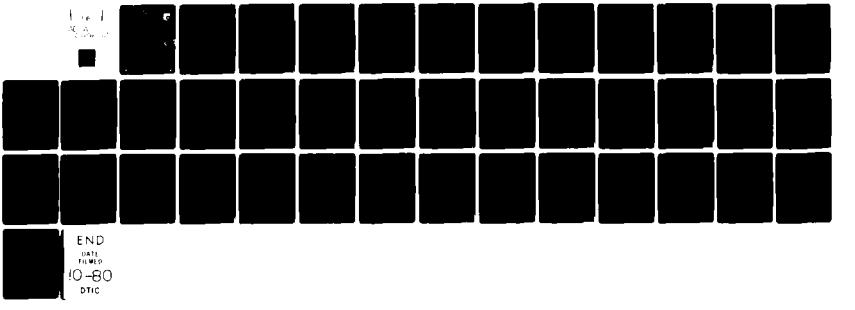

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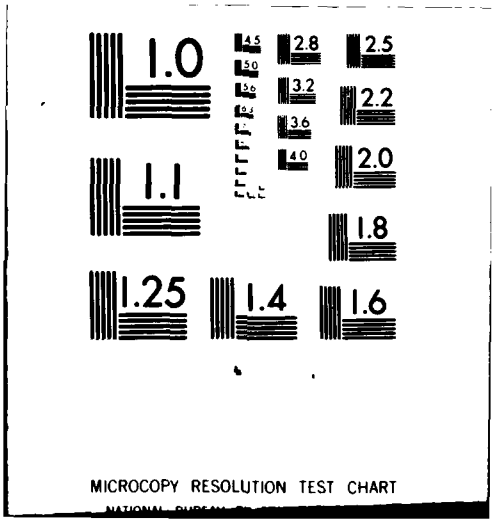
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SYNTHESIS OF MICROSTRUCTURES AND THE RELATIONSHIP BETWEEN MICROSTRUCTURE AND PROPERTIES

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University of Cincinnati

March 1980

TECHNICAL REPORT AFWAL-TR-80-4035

Final Report for period 15 May 1976 through 31 August 1979

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Lawrence R. Bidwell, Program Manager
Structural Metals Branch
Metals and Ceramics Division

FOR THE COMMANDER

Gail E. Eichelman, Chief
Structural Metals Branch
Metals and Ceramics Division

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The interrelationships of microstructure, processing and mechanical properties were investigated in several materials. In titanium alloy systems fatigue behavior, high temperature behavior, properties of powder compacts, and the effects of additions of hydrogen were studied. The aluminum research was concerned primarily with powder products and their unique properties. Other areas of investigation included quantitative metallography and fatigue behavior of AF 1410.

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FOREWORD

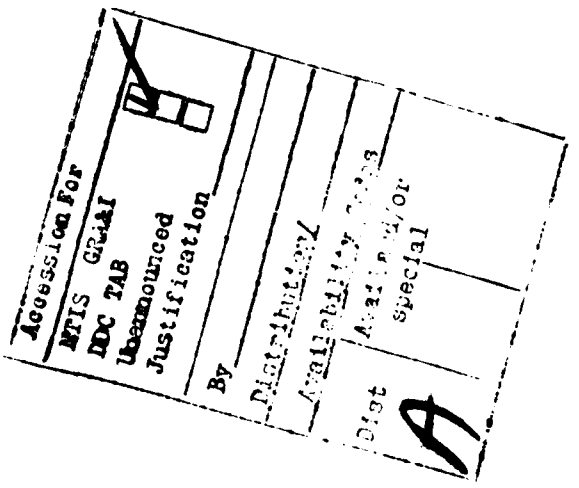
This Final Technical Report describes the work performed under Contract F33615-76-C-5227 entitled, "Exploratory Development of the Synthesis and the Related Microstructures and Properties of Material," for the period May 1976 - August 1979. The program is documented under Project 7351.

All contractual work for this Program was performed by the Department of Materials Science and Metallurgical Engineering, University of Cincinnati, under the direction of Dr. L.R. Bidwell, Structural Metals Branch, Metals and Ceramics Division, Materials Laboratory, AFWAL/MLLS Air Force Wright Aeronautical Laboratories, Wright-Patterson Air Force Base, Ohio 45433.

This report has been prepared by Mark Rosenblum and Dr. Daniel Eylon under Dr. Michael Hoch, principal investigator.

The major subcontractor to this program was the INCO Research and Development Center, Sufferin N.Y. The program was under the direction of R.D. Schelleng, Senior Project Manager.

A portion of the research, The Fatigue Crack Propagation and Corrosion Fatigue of AF 1410 Steel was performed at the University of Cincinnati by Prof. Stephen D. Antolovich.



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SECTION I

INTRODUCTION

This program was directed towards establishing relationships between microstructure and mechanical properties of materials. Specifically, titanium alloys, aluminum alloys, and a steel were studied utilizing most of the techniques available to researchers. A portion of the program was devoted to the development of new investigative tools. When expertise was required in other areas arrangements were made to include outside sources for portions of the program. Since most of the research has appeared in the public domain in other forms, this report contains summaries of the investigations and a reference list.

SECTION II

MICROSTRUCTURE MECHANICAL PROPERTY CORRELATIONS IN TITANIUM ALLOYS

The titanium alloy work under this topic is divided into three areas:

a) Fatigue behavior of Ti alloys

b) High temperature behavior of Ti alloys

c) Mechanical properties of Ti alloy powder compacts.

Each of these areas is reported here separately. Research subjects, like High Temperature Low Cycle Fatigue, which can be included in two areas (in this case, fatigue and high temperature), will be reported only in one.

In previous work carried out on Contract F33615-73-C-5097 it was demonstrated that the microstructure of Ti alloys can greatly influence properties like notch fatigue [1], smooth fatigue [2], high temperature notch fatigue [2], tensile strength and ductility [3], fatigue crack propagation [2,4] and creep [2]. In the case of titanium alloy powder compacts, the fatigue properties were not only related to microstructure, but also to the level of contaminants in the powder [5]. To enable an indepth study of defect related failures, the precision sectioning method [6] was developed and premature failures of IMI-685 fan disk material were investigated [7]. The three major research activities stemmed from this work.

1. FATIGUE BEHAVIOR OF TITANIUM ALLOYS

This work is divided into the following subareas:

a) Effect of defects on fatigue life of titanium castings.

b) Effect of HIP on fatigue life of titanium castings.

c) Fatigue crack propagation under dwell conditions.

d) Fatigue crack propagation in large colony beta annealed microstructures.

e) Fatigue mechanisms in beta annealed titanium alloys.

f) High temperature low cycle fatigue of near alpha titanium alloys.

a. Effect of Defects on Titanium Alloy Castings

The releationship between Ti-6Al-4V casting defects (mainly porosity) and high cycle room temperature smooth axial fatigue was investigated. Defects were classified into three major categories [8]:

dendritic pores
shrinkage pores
gas pores.

Dendritic pores were associated with the largest fatigue life debit [9,10,11]. Gas pores, due to their spherical shape, did not cause substantial fatigue life degradation. In many cases, lower fatigue lives were associated with large colony structure or massive grain boundary alpha phase [11].

Figure 1 is a data summary plot of Ti-6Al-4V smooth bar fatigue results in the annealed condition. It shows the relatively low fatigue strength of cast material, when compared to wrought material and powder compacts.

b. Effect of HIP on the Fatigue Properties of Titanium Castings

In an attempt to improve the poor fatigue properties of castings, some pore containing castings were HIP'd, resulting in complete closure of all pores. Fatigue test results [12] showed that the HIP'd material fatigue strength

is higher than that of as cast material but lower than wrought material. Precision sectioning analysis [6] showed that large colonies and massive grain boundary alpha phase were responsible for premature fatigue failures. A very similar mode of failure was also detected in Ti-17 powder compacts HIP'd above the beta transus temperature [13].

c. Fatigue Crack Propagation under Dwell Conditions

In previous work[7], a dwell effect that reduced the HCF life of beta processed IMI-685 was observed. In an attempt to understand the severe fatigue life loss the effect of dwell on the FCGR of various alloys was investigated. The work included [14]:

Ti-6Al-4V cross rolled plate (3 microstructural variations)
Ti-6Al-4V highly textured plate (standard and ELI grade)
Ti-11-beta forged
IMI-685-beta forged.

None of the tested materials showed any FCGR acceleration in the ΔK range of 12-25 MPa$\sqrt{m}$ under dwell conditions. On the contrary, some conditions [14] showed lower crack growth rates under dwell due to development of a highly tortuous crack path [15,16,17].

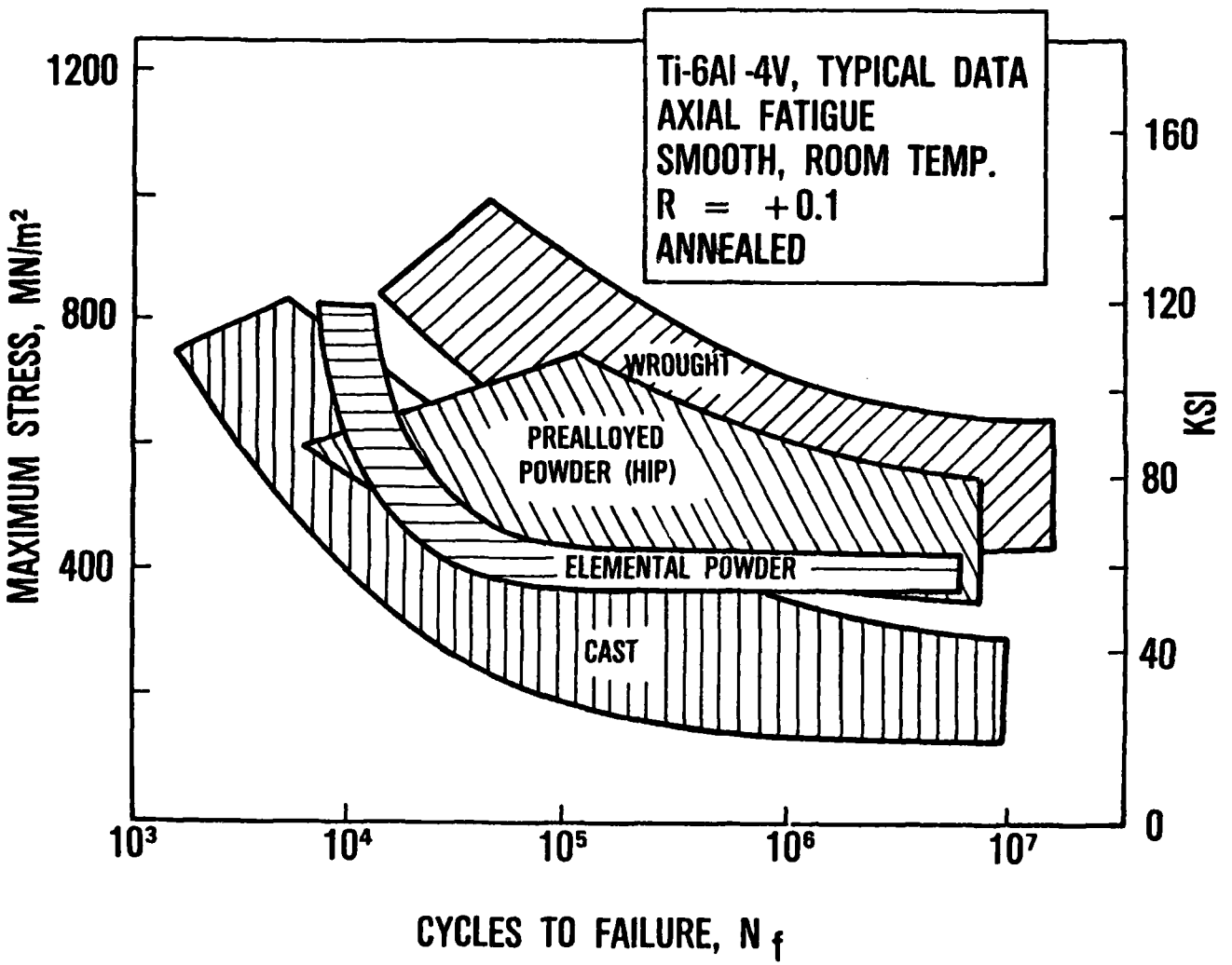

Figure 1. Room temperature smooth bar axial load (R=+0.1) fatigue results of cast, powder compact and wrought Ti-6Al-4V in annealed condition.

d. Fatigue Crack Propagation in Large Colony Beta Annealed Microstructures

To establish the mode of cracking associated with the development of alpha plate colonies of similarly aligned grains, large colonies were grown in thin Ti-11 FCG specimens in a way that only two colonies were across the specimen thickness [18]. The measured FCGR within individual colonies was similar to the growth rates in $\alpha+\beta$ anneal structures. However, due to crack branching, high crack tortuousity and crack renucleation at colony boundaries, the overall FCGR in this microstructure was slower than in $\alpha+\beta$ annealed structures. In beta annealed alloys containing many colonies across the specimen thickness, the FCGR was found to be lower with increasing colony size [19]. It should be noted, as a warning, that although large colony microstructures possess good FCG properties, they are extremely unstable for crack initiation [7] due to multiple cracking [4,18], leading to lower fatigue life and a large scatter of the results [20].

Similar behavior was also observed in banded $\alpha+\beta$ worked and annealed microstructures [21,22] which developed a high level of texturing and high aspect ratio alpha grains as a result of unidirectional hot work. These structures also provided crack branching and faceted crystallographic fatigue cracks.

e. Fatigue Mechanisms in Beta Annealed Titanium Alloys

The fatigue behavior of beta annealed titanium alloys is controlled by intense shear band formation from the initiation stage up to overload fracture. This slip behavior was found to be on both basal and prismatic planes of the hexagonal alpha phase structure [23,24]. The intense shear is generated both in the monotonic [23] and cyclic loading [24]

modes. Shear related crack initiation was observed in beta processed alloys like Ti-11 [2], IMI-685 [7], and Ti-6Al-4V [25] as well as in Ti-6Al-4V castings [8-12]. Shear band formation also promoted subsurface crack initiation [7,26] which is typical of large colony microstructures. The crack propagation stage was also related to intense shear band formation by inducing branch cracks following intense shear bands [18]. Even overload failure was found to be associated with the intense shear bands [2,18], but at the higher ΔK levels more bands were generated and their influence seems to be reduced due to the more homogeneous slip.

Using the same investigation techniques, fatigue crack mechanisms of experimental Al-7X75 alloys [27] were investigated. The highly textured material showed {111} shear band formations responsible for fatigue crack initiation, propagation and overload fracture [28].

It was also found in these beta annealed titanium alloy microstructues that faceted shear related fracture can be extended over several colonies [29]. Due to the Burgers' relationship, alpha plate colonies, within the same beta grain, can share a common basal plane orientation and therefore develop basal slip throughout several colonies. This type of multicolony faceted fracture was observed in initiation [30,31] and propagation [29] facets of wrought material and also in castings [12].

The nature of initiation and propagation facets was further investigated by the selected area electron channeling technique [31]. It was found that pure basal orientations exist only in nondwell cycling conditions, while dwell cycles promoted cracking on planes approximately 10 degrees from the (0001) orientation.

f. High Temperature Low Cycle Fatigue of Near Alpha Titanium Alloys

The strain control high temperature low cycle fatigue of Ti-5522S and Ti-6242S was investigated [32] in the temperature range of 800-1000°F. It was found that the HTLCF life decreases with increasing test temperature and decreasing test frequency (longer exposure times) [33]. The α+β processed microstructures showed much better HTLCF strength than the β processed material. Precision sectioning of crack initiation sites showed [34] that the long alpha plate interfaces are responsible for the lower fatigue strength of the β-annealed material. The dwell effect on the HTLCF was also studied and was shown to be detrimental to fatigue strength [35] especially in β processed material.

2. HIGH TEMPERATURE BEHAVIOR OF TITANIUM ALLOYS

The high temperature work is divided into the following subareas:

a) High temperature low cycle fatigue (reported in a previous section)

b) Creep properties of titanium alloys

c) Ion plating effects on high temperature properties.

a. Creep Properties of Titanium Alloys

Previous work [2], showed a strong relationship between the microstructure and creep strength of near alpha titanium alloys. In this work the effect of texture [36] on creep behavior was investigated as well as the effect of the environment [37]. It was shown that the creep resistance is greatest when the creep load direction is perpendicular to the basal planes [36], and when the creep is performed in vacua rather than in air [37]. The environmental effect work was later extended to include the ion plating work.

b. Ion Plating Effects on High Temperature Properties

The work described in the previous section demonstrated that in the absence of oxygen, the creep strength of $\alpha+\beta$ titanium alloys is increased. This led to experiments in protecting the alloy surface from the atmosphere by means of coatings. After several coating methods had been examined [39], ion plating with nobel metals like Au or Pt was found to be the most satisfactory in terms of stability and property improvement. This work was divided into the following sections:

a) Creep property investigation

b) Fatigue property investigation

c) Friction and wear

d) Component evaluation.

c. Creep Properties of Ion Plated Titanium Alloys

Both Au and Pt coatings were tested [39]. The Pt coat showed better high temperature stability [40,41] due to lower solubility in titanium at higher temperatures. The ion plated material showed better creep resistance than uncoated material in air but was inferior in creep in vacuum [42]. The coating effect was more pronounced in high surface to volume ratio specimens [43] and did not exist in alloys containing only alpha phase. It was then concluded that creep of uncoated specimens in air, allows oxygen to diffuse along the α/β interfaces relatively deep into the material and transforms some of the β phase into α [42]. The oxygen induced transformation increases the mobile dislocation density and hence the creep rate [45].

d. Fatigue Properties of Ion Plated Titanium Alloys

The smooth high cycle fatigue life of Pt ion plated Ti-6242S was greatly improved when compared to uncoated material [46,47]. The fatigue strength improvement was higher at $800^{\circ}F$ than at room temperature [48]. Crack initiation analysis showed that fatigue cracks of coated material tested at $800^{\circ}F$ were from subsurface locations [46,49]. This indicated surface initiation suppression either due to implantation surface hardening [47] or due to the prevention of surface oxide layer formation [50].

e. Component Evaluation

The possible applications of this technology are discussed in detail in a joint Navy/Air Force report [53].

f. Patents

As a result of this research activity, two patent applications were successfully filed by S. Fujishiro and D. Eylon, and patents #4,137,370 [51] and #4,181,590 [52] were granted.

3. MECHANICAL PROPERTIES OF Ti ALLOY POWDER COMPACTS

This area is divided into the following sections:

a) Crack Initiation analysis of powder compacts

b) Characterization of loose powders

c) Evaluation of prealloyed powder compacts

d) Evaluation of elemental powder compacts.

a. Crack Initiation Analysis of Powder Compacts

This work was the extension of the work reported in a previous section. By using the precision sectioning technique [6], premature fatigue failures or powder compacts originating from subsurface locations were analyzed. This was not limited only to titanium alloy compacts [55], but also included superalloy [56] and aluminum alloy compacts [57].

In most cases, the defect initiation sites were at nonmetallic inclusions. They were divided into three types:

a) Deep subsurface initiation

b) Near surface initiation

c) Surface initiation

The deep subsurface initiation was observed in cases where only a few inclusions existed in the gage volume [5,55], and they were of a substantial size (larger than the average grain size). The titanium alloy powder compact fatigue behavior is reported in detail elsewhere [58].

b. Characterization of Loose Powders

Loose powder characterization [59] was performed on the powders listed in Table 1. This work covered almost all the powders available today on a commercial or an experimental basis. The morphological, microstructural, and defect characterization work is described in detail in other reports [60,61].

c. Evaluation of Prealloyed Powder Compacts

This part of the work started near the end of the contract. Most of the effort involved preparation and processing of Ti-6Al-4V REP powder for consolidation by the HIP and SEP+HIP methods. These consolidation methods are discussed in detail in other reports [62,63].

d. Evaluation of Elemental Powder Compacts

In this part of the work, cold pressed and sintered Ti-6Al-4V powder compacts (prepared by cold pressing and sintering Ti sponge fines mixed with Al + V master alloy powder) were tested for tensile fracture toughness and high cycle fatigue strength [64]. Although the compacted material contained one vol. pct. porosity, the tensile strength and ductility were surprisingly high. On the other hand, the high cycle fatigue strength (Figure 1) was low relative to HIP'd prealloyed powders and wrought material [65].

Table 1 Powder Making Methods and Alloys Characterized

Powder Making Method	Designation	Basic Particle Shape	Developing Organization/ Current Producer	Alloys Characterized
Sponge Fines (Na reduction, Elemental)		Angular	RMI/Gould/Dynamet(USA)	CP titanium, Ti-6-4
Sponge Fines (electrolytic), Elemental		Angular	Howmet/Gould (USA)	CP titanium, Ti-6-4
Hydride-Dehydride	HDH	Angular	Oremet (USA)	Ti-6-4
Rotating Electrode Process	REP	Spherical	NMI (USA)	Ti-6-4, Ti-6-2-4-6, Ti-17 Corona-5, Beta-III
Plasma Rotating Electrode Process	PREP	Spherical	NMI (USA)	Ti-6-4
Hydrogen Evolution (EB)	C-T	Generally Spherical	Crucible (USA)	Ti-6-4, Ti-17
Electron Beam Rotating Disc	EBRD	Partially Spherical	Leybold-Hereaus (W. Germany)	Ti-6-4
Powder Under Vacuum (EB)	PSV	Spherical	CENG (France)	Ti-6-4, IMI-829
Pendant Drop (EB)	PD	Elliptical	Battelle/GTE (USA)	Ti-6-4
Strain Energizing Process	SEP	Disc	Kelsey-Hayes (USA)	Ti-6-4

SECTION III

HYDROVAC

Earlier research at AFWAL/ML with titanium wrought alloys [66] and powder into which hydrogen had been introduced demonstrated a reduction in flow stress during compaction. Others had also noted a similar effect in the forging of wrought materials [67,68]. The advantages offered by such a process led to an investigation into the effects of hydrogen on titanium alloys. In these studies wrought material, primarily Ti-6Al-4V, was used to establish a basis for application of the hydrovac process [69].

Once the procedures for reproducibly hydrogenating titanium were established, an initial forging study demonstrated that 0.4 wt% hydrogen addition would lower flow stress by 30%. In this effort, the hydrogen was retained by oxide and glass coating. Work was also initiated to determine phase relationships and a beta transus curve established.

The effects of the addition and removal of hydrogen were studied. Hydrogenated samples, 0.5% hydrogen, were used to attempt to establish the necessary parameters for subsequent removal. It was found that the vacuum treatment time for complete hydrogen removal in the range 650°C to 815°C was longer than the time required in a diffusion controlled process. Microstructures produced by the addition and removal of hydrogen resembled the starting structure except for some formation of grain boundary alpha phase at the higher temperature.

Further work on phase equilibrium revealed that the transus curve developed might actually represent the presence of a eutectoid temperature at approximately 800°C. Eutectoid decomposition, when allowed to proceed to completion produced an extremely fine microstructure consisting of alpha, beta, and hydride phases [70]. This modification of microstructure gave rise to ideas about mechanical property modification through microstructural manipulation.

A tensile test program established that yield strength could be improved while retaining reasonable ductility. Continued work in this area will be necessary to establish the optimum Hydrovac treatment cycle. Heat treatment of dehydrogenated Ti-6Al-4V established that the microstructures observed could be further modified to suit a particular purpose. Some work was also carried out in other alloys, Ti-8Al-1Mo-1V and Ti-5Al-2.5Sn to determine the feasibility of applying Hydrovac.

SECTION IV

MICROSTRUCTURE-PROCESSING-MECHANICAL RELATIONSHIP IN ALUMINUM ALLOYS

The emphasis in the aluminum research program has been on powder metallurgy which can provide, through alloy modifications not possible with ingot products, superior mechanical property and temperature capabilities. Through a combination of research and subcontract studies the effects of powder processing and new methods of powder manufacture have been investigated. At the same time research has been completed in some aspects of more traditional alloy products.

Fatigue behavior and failure mechanisms of 7075 aluminum alloys modified to study the effects or purity level and dispersoid type on the fatigue behavior of 7000 series alloys were investigated. [27,28] Ten different compositions based on the 7075 alloy were produced with five levels of Fe+Si and either Cr or Zr dispersoids. Notched axial fatigue specimens were tested at room temperature. The fatigue life did not correlate with either purity level or dispersoid type. Specimens failed by three macroscopic modes designated as: slant, vee, or flat fracture. Sectioning analysis showed that the slant, vee, and flat fractures resulted from single, double and multiple initiation, respectively. Both initiation and propagation in all three modes of failure were dominated by slip related fracture on {111} planes inclined at 35° to the tensile axis of the textured material. The same failure mechanisms were observed in smooth fatigue specimens.

As an introductory study in powder aluminum alloys, research was initiated to examine the effect of varying quench rates from the solution treatment temperature and the effect of varying heating rates to final aging temperature on the hardness of MA87, a powder aluminum alloy of current

interest to the Air Force [71]. The results obtained were compared with results of similar research on the ingot metallurgy alloy 7075 and were supplemented with electron microscopy. From a hardness standpoint, MA87 and 7075 have similar quench sensitivity behavior. The quench sensitivity in MA87 appears to be due to solute depletion in the matrix through two principal means, grain boundary precipitation and preferential nucleation at Co_2Al_9 and oxide particles in the matrix. MA87 exhibits a reversion response after natural aging. Natural aging appears to provide no advantages for MA87 aged at ($120^{o}C$).

Early indications showed that certain combinations of forging procedures and compositions demonstrated superior fatigue crack growth rates in MA87. [72] After a great deal of work to characterize the forged material and determine differences in microstructure at all levels, optical to submicroscopic, it was determined that their specific property improvements had been caused by improper treatment and were not reproducible. Consequently, the studies were directed toward establishing the processing parameters which would improve mechanical properties through changes in microstructure. During this research, several new techniques were applied. The da/dN ring sample was used to determine fatigue crack growth rates [73]. With this sample configuration the load can be calculated from the equation:

$$\frac{KB\sqrt{Ro}}{P} = 2.61$$

Where K is the stress intensity factor, B is the ring thickness, Ro is the outside ring radius, and P is the applied load. Transmission electron microscope specimens were obtained by breaking the cracked rings, cutting just below the fracture surface, grinding the fracture surface until evidence of fracture disappears and then grinding the

other side. Areas of interest are cut out and electropolished to remove grinding damage than polished until perforation.

It became evident that the potential of powder aluminum alloys could be extended through the application of different powder production techniques. Consequently, the University of Cincinnati entered into a subcontract arrangement with the International Nickel Company to gain an understanding of the factors controlling room and elevated temperature strength in aluminum alloys prepared by mechanical alloying. [74] Mechanical alloying is a solid-state powder milling process which produces oxide dispersion strengthened aluminum powders. The work conducted focused on two compositional aspects. First, a study of the relative effectiveness of oxide or carbide dispersoids was carried out and, second, the potential roles of additions of several metallic alloying elements singly and in combination were assessed. In addition to these aspects, the possibility of further strengthening through grain size control via variations in thermomechanical processing was studied. The primary goals established for the work included increased tensile and creep strength at 450 and 650^{o}F, with consideration given to fatigue, toughness and corrosion resistance.

The grain size in these materials, stabilized by the presence of oxide and carbide dispersoid particles, results in a strength at room temperature approximately equal to that of EC Al hardened to the H19 temper. The fine grain size is produced by the severe cold working of the powder during mechanical alloying. The dispersoid stabilizes this grain size and, if present in sufficient quantity, renders the material virtually immune to grain growth. The oxygen and carbon contents establish the quantity of dispersoid which forms. In addition

to grain size control, the dispersoid contributes to strength, apparently by impeding dislocation movement. The stabilizing effect of the dispersoid particles on grain growth is further demonstrated by the absence of room temperature strength loss following 100-hour exposures at temperatures as high as 650°F.

Cu and Mg additions were found to exert a profound influence on strength. Unfortunately, these elements and Zn severely decreased elevated temperature properties. Additions of the low solubility elements, Fe, Ni, Co, Ti, Mn and Cr were found to increase strength. However, the strengthening increment was essentially independent of element selected.

Tensile properties at 450°F also depended on dispersoid and low solubility element contents, but to a lesser extent than at room temperature. At 550 and 650°F, tensile properties continued to be slightly influenced by low solubility element content but they were independet of changes in dispersoid content.

Time-dependent mechanical properties were also influenced by alloying additions. Mg, Cu and Zn, adversely affected properties while the low solubility additions did not harm, and in some cases improved, strengths. The alloy additions most beneficial to stress rupture properties were Ti or a combination of Fe and Cr. Combinations of Cr and Ti or Fe and Ti did not improve properties beyond those exhibited by the unalloyed materials. Creep response, evaluated only for the ternary alloys, appeared to be independent of alloying element addition. This work is reported in greater detail elsewhere [74].

SECTION V

QUANTITATIVE METALLOGRAPHY

Initial studies [75] in microstructural analysis of Ti-6242 with computers and other techniques failed to establish definitive relationships to mechanical properties. Two approaches, very specific measurement of each microstructural feature and attempts to recognize some pattern in the microstructure were employed [76,77]. In neither case were metallurgists or people familar with established techniques for microstructural analysis involved in the basic feature selection decisions. The failure to obtain significant levels of correlation between the measured microstrucutre and the accompanying set of mechanical properties was partly laid to this lack of specific expertise. Upon further examination it was found that the supposedly sophisticated computer software was not performing as expected and the microstructural images were not analyzed correctly. Further work to improve the computer performance was not satisfactory and was discontinued.

In order to establish validity of the basic approach, mechanical properties of a titanium alloy can be predicted from measurements of the microstructure (alpha phase), the classical "hand measurement" techniques of quantitative metallography were employed. If successful, the microstructural parameters of importance could then be related to the quantities used in the original computer analysis. The primary and secondary alpha phases were measured separately to aid in final analysis.

The quantities measured were:

P_p - fraction of test grid points inside features

N_A - number of features per unit area

N_L - number of features per unit line length.

The relation between these measured quantities and the geometric properties measured by the computer are:

Area Fraction - P_p

Mean Particle Size - $\bar{A}$

Mean Free Path - λ

Mean Interval Length - σ

Number of Intervals - N_L

Mean Duration - $\bar{L}$

$\bar{A} = P_p/N_A$

$\lambda = (1-P_p)/N_L$

$\sigma = 1/N_L$

$\bar{L} = P_p/N_L$

Statistical validity was provided by using 25 "throws" of the grids as each microstructure for both primary and secondary alpha measurements. The order of picture examination was preserved and a second set of measurements taken to detect the development of counter bias. Actual specimens were measured in a metallograph to demonstrate "typical microstructure" accuracy of the original micrographs.

The data is presented in Tables II to IV for primaries, secondaries, and combined. The combined data represents a stereological combination of the primary and secondary data. Mechanical property and original specimen data are presented in Table V. Linear correlations, geometric properties vs geometric properties and geometric properties vs mechanical properties are also tabulated, Tables VI, VII.

The correlations between microstructure and properties were still not highly significant. Additional attempts to establish nonlinear correlations were also unsucessful. At this point it became necessary to reexamine the

TABLE II

GEOMETRIC PROPERTIES - PRIMARIES

	Measured Quantities			Calculated Quantitites			
	A_A	N_A	N_L	A_i	σ	L_2	λ
1	0.256	5.67	1.38	0.0545	0.725	0.224	0.501
2	0.310	6.41	1.35	0.048	0.742	0.230	0.512
3	0.123	1.46	0.35	0.084	2.889	0.355	2.534
4	0.268	8.16	1.45	0.033	0.688	0.184	0.503
5	0.450	23.40	3.28	0.019	0.305	0.137	0.168
6	0.431	11.43	2.13	0.038	0.469	0.202	0.267
7	0.473	19.23	2.87	0.025	0.349	0.165	0.184
8	0.335	5.76	1.30	0.058	0.770	0.258	0.512
9	0.358	7.51	1.52	0.048	0.657	0.235	0.422
10	0.388	8.03	1.68	0.048	0.596	0.231	0.365
11	0.333	9.54	1.67	0.035	0.597	0.199	0.399
12	0.423	7.59	1.65	0.056	0.605	0.256	0.349
16	0.355	5.04	1.36	0.070	0.733	0.260	0.473
17	0.409	6.00	1.45	0.068	0.692	0.283	0.409
18	0.323	10.88	2.17	0.030	0.461	0.149	0.312
19	0.320	6.69	1.35	0.048	0.743	0.238	0.505
20	0.310	8.57	1.44	0.036	0.693	0.215	0.478
21	0.428	15.63	2.54	0.027	0.393	0.169	0.225
22	0.380	6.55	1.67	0.058	0.599	0.228	0.371
23	0.278	6.61	1.22	0.042	0.822	0.229	0.594
24	0.429	10.26	2.07	0.042	0.483	0.207	0.276
25	0.275	8.49	1.91	0.032	0.525	0.144	0.380
26	0.335	14.57	2.60	0.023	0.384	0.129	0.248
27	0.531	17.80	3.13	0.030	0.319	0.170	0.150
28	0.489	15.19	2.73	0.032	0.366	0.178	0.187
32	0.410	15.35	2.53	0.027	0.395	0.162	0.233
33	0.463	14.45	2.58	0.032	0.387	0.181	0.206
34	0.386	13.48	2.57	0.027	0.389	0.150	0.239
35	0.105	2.88	0.54	0.037	1.862	0.196	1.667
36	0.120	3.54	0.62	0.034	1.603	0.192	1.410
37	0.193	7.52	1.10	0.026	0.912	0.176	0.736
41	0.173	6.07	1.05	0.029	0.952	0.165	0.788
42	0.110	3.54	0.48	0.031	2.096	0.231	1.866
43	0.083	4.90	0.61	0.017	1.634	0.136	1.498
44	0.298	5.81	1.45	0.051	0.690	0.206	0.485
45	0.385	7.25	1.59	0.053	0.628	0.242	0.386
46	0.353	6.99	2.02	0.051	0.495	0.175	0.320
47	0.348	8.91	1.73	0.039	0.580	0.202	0.378
48	0.354	9.12	1.88	0.039	0.533	0.189	0.345
49	0.291	6.42	1.19	0.045	0.842	0.245	0.597
50	0.214	4.45	0.98	0.048	1.020	0.218	0.802
51	0.251	7.37	1.71	0.034	0.586	0.149	0.439
52	0.328	8.17	1.58	0.040	0.633	0.208	0.425

TABLE III

GEOMETRIC PROPERTIES - SECONDARIES

	Measured Quantities			Calculated Quantities		
	A_A	N_A	N_L	A_i	L_2	λ
1	.095	13.27	1.28	0.007	0.074	0.707
2	.050	10.92	1.20	0.005	0.042	0.791
3	.539	24.02	4.91	0.022	0.110	0.093
4	.070	6.28	0.94	0.011	0.075	0.989
5	.003	2.77	0.49	0.001	0.006	2.039
6	.075	4.89	0.80	0.015	0.094	1.161
7						
8						
9						
10						
11						
12						
16						
17						
18						
19						
20						
21						
22						
23	.301	42.20	5.28	0.007	0.057	0.132
24	.056	4.83	0.55	0.012	0.103	1.732
25	.045	2.28	0.47	0.020	0.096	2.024
26	.070	32.44	2.85	0.002	0.025	0.327
27						
28						
32	.126	13.16	1.73	0.010	0.073	0.504
33	.066	10.09	1.29	0.007	0.051	0.724
34	.076	8.12	1.25	0.009	0.061	0.737
35	.491	51.75	7.57	0.010	0.065	0.067
36	.300	60.28	6.64	0.005	0.045	0.106
37						
41	.688	50.95	7.18	0.014	0.096	0.043
42	.623	19.72	5.09	0.032	0.123	0.074
43						
44	.118	3.66	0.93	0.032	0.127	0.946
45	.126	8.88	1.51	0.014	0.083	0.578
46	.110	7.19	1.25	0.015	0.088	0.711
47	.143	16.37	2.12	0.009	0.067	0.404
48	.128	8.613	1.62	0.016	0.079	0.543
49	.199	14.72	2.40	0.014	0.083	0.334
50	.323	28.24	4.14	0.011	0.078	0.163
51	.283	11.54	3.39	0.025	0.084	0.212
52	.200	13.23	2.08	0.015	2.096	0.386

TABLE IV

GEOMETRIC PROPERTIES - COMBINED

	Measured Quantities			Calculated Quantities			
	A_A	N_A	N_L	A_i	L_2	λ	HEAT O_2
1	.351	18.94	2.66	0.019	0.132	0.243	0.13
2	.360	17.33	2.55	0.021	0.141	0.251	0.13
3	.662	25.48	5.26	0.026	0.126	0.064	0.10
4	.338	14.44	2.39	0.023	0.141	0.277	0.10
5	.453	26.17	3.77	0.017	0.120	0.145	0.10
6	.506	16.32	2.93	0.031	0.173	0.169	0.10
7	.473	19.23	2.87	0.025	0.165	0.184	0.10
8	.335	5.76	1.30	0.058	0.258	0.258	0.10
9	.358	7.51	1.52	0.048	0.236	0.422	0.10
10	.388	8.03	1.68	0.048	0.231	0.364	0.11
11	.333	9.54	1.68	0.035	0.199	0.399	0.10
12	.423	7.59	1.65	0.056	0.256	0.350	0.14
16	.355	5.04	1.36	0.070	0.261	0.474	0.13
17	.409	6.00	1.45	0.068	0.282	0.408	0.12
18	.323	10.88	2.17	0.030	0.149	0.312	0.11
19	.320	6.69	1.35	0.048	0.237	0.504	0.09
20	.310	8.57	1.44	0.036	0.215	0.479	0.12
21	.428	15.63	2.54	0.027	0.169	0.225	0.12
22	.380	6.55	1.67	0.058	0.228	0.371	0.14
23	.579	48.81	6.05	0.012	0.096	0.070	0.11
24	.485	15.09	2.62	0.032	0.185	0.197	0.11
25	.320	10.77	2.38	0.020	0.135	0.286	0.11
26	.405	47.01	5.45	0.009	0.074	0.109	0.11
27	.531	17.80	3.13	0.030	0.170	0.150	0.11
28	.489	15.19	2.73	0.032	0.179	0.187	0.11
32	.536	28.51	4.26	0.019	0.126	0.109	0.14
33	.529	24.54	3.87	0.022	0.137	0.122	0.14
34	.462	21.60	3.82	0.021	0.121	0.141	0.14
35	.596	54.63	8.11	0.011	0.074	0.050	0.14
36	.420	62.82	7.26	0.007	0.058	0.080	0.14
37	.193	7.52	1.10	0.026	0.176	0.734	0.14
41	.861	57.02	8.23	0.015	0.105	0.017	0.14
42	.733	23.26	5.57	0.032	0.132	0.048	0.14
43	.083	4.90	0.61	0.017	0.136	1.503	0.14
44	.416	9.47	2.38	0.044	0.175	0.245	0.14
45	.511	16.13	3.10	0.032	0.165	0.158	0.14
46	.463	14.18	3.27	0.032	0.142	0.164	0.14
47	.491	25.28	3.85	0.020	0.128	0.132	0.14
48	.482	17.73	3.50	0.027	0.138	0.148	0.14
49	.490	21.14	3.59	0.023	0.137	0.142	0.14
50	.537	32.69	5.12	0.016	0.105	0.090	0.14
51	.534	18.91	5.10	0.028	0.105	0.091	0.14
52	.528	21.40	3.66	0.025	0.144	0.129	0.14

TABLE V

MECHANICAL PROPERTIES

		UTS	±	YS	±	%EL	±	%RA	±	K_{IC}	±	$\left(\frac{K_{IC}}{Y.S.}\right)$	LCF PP
A	1	181.9	0.4	167.5	0.6	13.2	0.1	33.5	3.3	31.2	0.5	347.0	13,000
B	2	180.2	0.15	166.5	0.5	13.4	0.2	38.2	4.1	30.6	0.6	337.8	11,000
19	3	177.0	0	158.5	0.5	11.1	0.8	21.5	1.5	52.8	1.1	1109.6	12,000
20	4	179.0	1.0	164.0	1.0	13.9	1.0	34.0	8.3	43.9	0.3	716.5	15,000
21	5	175.5	1.0	165.5	1.5	14.0	0	35.5	0	38.4	1.0	538.2	12,500
22	6	179.0	1.5	167.5	0.5	13.9	0.3	39.5	3.0	39.2	0.8	547.6	17,000
25	7	180.9	0.6	170.2	0.8	12.6	1.2	37.6	7.2	32.6	0.4	366.9	14,500
LKCN	8	179.8	0.2	169.3	0.8	11.5	0.5	31.9	1.4	28.2	0.4	277.5	13,000
2LKCN	9	178.1		167.2		13.5		36.0		32.3	1.8	372.9	16,900
LKWT	10	180.1	1.9	166.9	1.8	14.2	0.2	39.7	1.5	26.9	0.1	259.8	25,100
LLRB	11	178.7	2.3	169.5	2.0	11.5	0.5	31.0	0.6	31.9	0.8	354.2	18,800
LNJJ	12	190.5	3.0	180.2	3.1	11.8	0.8	32.3	0.5	28.4	0.7	248.4	32,400
PHIL-1	13	182.2	1.0	166.5	0.5	12.4	1.2	37.9	4.1	41.6	0.6	624.3	6,500
PHIL-2	14	174.6	0	164.0	1.0	13.3	0.3	35.7	0	39.8	0.6	583.9	8,500
SGLO	15	176.1	1.7	159.5	1.5	15.7	1.1	51.1	2.3	59.4	2.2	610.2	14,500
WCLX	16	201.8		181.8		8.5		25.9		27.4	1.7	227.2	16,500
WDAR	17	194.5	1.5	176.1	2.9	11.3	4.3	17.5	0.4	24.7	0.05	196.7	20,300
ZWLC	18	180.0	4.5	170.3	3.3	15.3	0.2	41.6	4.3	37.5	4.7	484.8	33,400
ZWLD	19	179.1	0.9	161.3	0.3	14.7	0.4	41.4	5.5	31.5	1.0	381.4	38,750
ZWNN	20	199.8	4.7	186	4.2	10.5	1.8	29.4	8.3	27.3	0.6	215.4	27,000
ZZWNN	21	190.2	1.7	182.6	1.6	11.3	2.3	28.8	3.2	27.4	1.65	225.2	39,800
ZWUY	22	188.4	2.0	175.3	3.1	8.8	0.3	17.6	0.9	24.2	0.1	190.6	54,400
110	23	170.0	0	161.0	1.0	15.8	0.3	31.0	2.5	37.4	5.8	539.6	20,200
120	24	168.5	1.5	161.0	1.0	14.8	0.3	37.5	3.5	29.0	0	324.5	10,400
130	25	182.5	0.5	172.5	0.5	11.5	1.0	30.0	2.5	25.4	0.3	217.1	12,700
210	26	170.0	0	162.5	0.5	12,3	1.3	34.3	3.3	31.0	1.4	363.9	16,700
220	27	169.0	3.0	164.0	2.0	10.8	2.3	25.0	9.0	27.0	2.0	271.1	18,600
230	28	173.5	0.5	168.5	0.5	11.5	1.5	27.5	5.5	24.9	0.5	218.2	20,200
310	29	170.5	0.5	162.5	0.5	12.5	1.0	31.3	2.3	32.2	0.8	392.7	22,900
320	30	171.5	0.5	164.0		12.3	1.3	28.5	3.5	30.2	0.7	339.1	13,500
330	31	180.5	0.5	172.0	0	9.3	0.8	17.5	3.5	27.8	1.0	261.2	26,100
410	32	178.0	1.0	171.5	0.5	12.5	0	36.3	0.3	22.7	0.3	175.2	36,900
420	33	182.0	1.0	175.0	0	12.8	0.3	36.3	0.8	20.9	1.1	142.6	32,700
430	34	188.5	1.5	180.5	1.5	11.8	0.3	31.8	1.8	18.9	.55	109.6	33,200
510	35	169.0	4.0	157.5	2.5	10.5	1.5	23.5	1.5	43.1	0.7	748.9	19,100
520	36	179.0	0	164.0	0	9.8	1.8	17.5	5.5	30.5	0.65	345.9	30,000
530	37	204.5	0.5	193.5	0.5	6.0	0.5	15.3	5.8	17.4	0.5	80.9	37,200
610	38	175.5	0.5	159.5	0.5	4.5	0.3	8.5	1.5	46.3	0.9	842.6	23,100
620	39	177.0	0	165.0	1.0	3.8	0.3	7.0	0	44.7	0.4	733.9	13,350
630	40	186.0				<1.0		<1		27.6	1.0		33,100
710	41	178.0	1.0	155.5	2.5	13.3	0.3	28.0	0	34.0	0.15	478.1	
720	42	180.0	0	165.5	0.5	11.8	0.8	29.5	0.5	31.5	1.7	362.3	32,150
730	43	196.5	3.5	180.0	2.0	7.5	0	22.5	2.5	24.1	0.2	179.3	73,900
810	44	179.5	0.5	168.5	0.5	12.5	0.5	26.5	0.5	28.2	0.3	280.1	16,100
820	45	180.0	0	169.0	0	10.3	1.8	27.0	1.0	28.0	0.6	274.5	25,400
830	46	194.0	0	180.0	0	8.5	0	18.0	2.0	23.1	0.1	164.7	26,400
910	47	176.5	1.5	166.5	1.5	12.0	2.0	26.5	4.5	29.5	0.15	313.9	22,800
920	48	181.5	0.5	170.0	0	12.3	0.8	28.5	1.5	27.0	0.5	252.3	27,800
930	49	189.0	1.0	174.0	0	8.5	0.5	13.5	0.5	25.9	0.05	221.6	24,500
1010	50	168.5	1.5	156.5	1.5	10.0	1.0	23.8	2.8	36.6	0.05	546.9	20,100
1020	51	170.5	0.5	159.5	0.5	12.5	2.0	27.0	0	30.8	0	372.9	26,600
1030	52	177.0	3.0	167.0	3.0	10.5	1.5	22.8	2.3	27.0	0.35	261.4	36,600

TABLE VI

CORRELATION MATRIXES - GEOMETRIC VS. GEOMETRIC

PRIMARIES

	N_A	N_L	N_i	α	L_2	λ
A_A	.74	.86	-.02	-.82	-.10	-.84
N_A		.94	-.57	-.66	-.57	-.64
N_L			-.41	-.80	-.53	-.79
A_i				.29	.89	.22
σ					.46	.997
L_2						.383

SECONDARIES

	N_A	N_L	N_i	α	L_2	λ
A_A	.68	.89	.38	-.67	.38	-.68
N_A		.93	-.25	-.68	-.20	-.66
N_L			.05	-.74	.04	-.74
A_i				-.06	.87	-.10
σ					-.05	1.0
L_2						-.10

COMBINED

	N_A	N_L	N_i	α	L_2	λ
A_A	.59	.78	-.27	-.79	-.39	-.79
N_A		.94	-.72	-.65	-.77	-.56
N_L			-.65	-.76	-.79	-.67
A_i				.45	.93	.30
σ					.58	.98
L_2						.43

TABLE VII

CORRELATION MATRIXES - GEOMETRIC vs. MECHANICAL PROPERTIES

PRIMARIES

	A_A	N_A	N_L	A_i	α	L_2	λ
UTS	-.04	-.16	-.15	.15	-.01	.11	-.02
YS	.21	.12	.15	-.05	-.25	-.08	-.26
EL	.28	.28	.30	-.10	-.27	-.27	-.28
RA	.37	.41	.42	-.23	-.37	-.21	-.37
K_{IC}	-.34	-.19	-.29	.19	.48	.27	.47
(K_{IC}/YS^2)	-.37	-.2	-.3	.21	.55	.29	.54

SECONDARIES

	A_A	N_A	N_L	A_i	α	L_2	λ
UTS	-.24	-.38	-.37	.13	.14	.15	.13
YS	-.58	-.57	-.62	-.01	.32	-.03	.32
EL	-.18	-.14	-.19	-.18	.34	-.17	.35
RA	-.42	-.37	-.45	-.28	.48	-.26	.49
K_{IC}	.43	.33	.42	.04	-.15	.03	-.15
(K_{IC}/YS^2)	.48	.35	.46	.05	-.19	.05	-.19
LCF	.19	.13	.19	.1	-.47	.08	-.47

COMBINED

	A_A	N_A	N_L	A_i	α	L_2	λ
UTS	-.49	-.50	-.53	.41	.58	.43	.55
YS	-.55	-.55	-.60	.34	.52	.41	.49
EL	.23	.14	.12	-.09	-.32	-.03	-.35
RA	-.05	-.10	-.16	-.02	-.04	.12	-.07
K_{IC}							
(K_{IC}/YS^2)	.15	.27	.21	-.22	-.24	-.19	-.23
LCF	-.28	-.15	-.17	-.01	-.49	.02	.54

basic approach. Did alpha phase morphology (optically resolvable) indeed determine mechanical properties? In light of more recent titanium work it must be concluded that the submicroscopic structure is as important as the optically visible structure and correlations could only be possible on a far more limited basis.

The grain growth of Ti-6Al-4V at superplastic forming temperatures was studied to determine the effectiveness of yttria additions as a growth inhibitor and to evaluate the commonly used measurement techniques [78]. Area fraction and mean linear intercept measurements were used. The mean linear intercept was obtained for all types of boundaries in the two phase structure. It was shown that mean linear intercept (commonly referred to as grain size) measurements of separate alpha and beta grain sizes did not differ in grain growth behavior from the standard all intercept grain size. It was also shown that yttria addition did indeed retard grain growth.

SECTION VI

FATIGUE CRACK PROPAGATION AND CORROSION FATIGURE OF AF 1410 STEEL

Extensive work was done on the fatigue of AF 1410 steel [79]. Experimental work included fatigue crack propagation, (FCP), corrosion fatigue (CF), and low cycle fatigue (LCF). The FCP studies showed a frequency effect at low stress intensities such that the lower the frequency the more rapid the crack growth per cycle. This was interpreted as being due to an interaction with the environment. When testing was done in dry Ar this effect disappeared. The FCP rate of this steel was not significantly superior to that of other high strength steels in its class. The data are well-represented by the ARCTA NH program which was developed under another Air Force contract.

The mechanisms of crack propagation were studied by scanning microscopy (SEM) and details of the deformation process were studied by transmission electron microscopy. The SEM studies showed intergranular propagation in either 3 1/2% NaCl or in air at low temperatures. The TEM results revealed a heavily dislocated martensite lath substructure and fatiguing had the effect of arranging the dislocations into very tight cells.

The LCF behavior of this steel showed an extraordinarily high slope (-1.13) when plotted in terms of a Coffin-Manson plot. This indicates that while the steel is quite good for high strain applications the low strain life is likely to be shorter than for other steels. The LCF results were used to make a theoretical prediction of FCP behavior using a model developed by S. Antolovich and some of his students. The results of this calculation are in excellent agreement with experiment.

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